

(19)



Europäisches Patentamt

European Patent Office

Office européen des brevets



(11)

EP 0 787 813 A1

(12)

## EUROPEAN PATENT APPLICATION

(43) Date of publication:  
06.08.1997 Bulletin 1997/32

(51) Int Cl.<sup>6</sup>: C22C 38/22, C22C 38/24,  
C22C 38/26, C22C 38/28,  
C22C 38/32

(21) Application number: 97101998.9

(22) Date of filing: 07.02.1997

(84) Designated Contracting States:  
DE FR GB

(30) Priority: 10.02.1996 JP 48130/96

(71) Applicants:

- SUMITOMO METAL INDUSTRIES, LTD.  
Osaka-shi, Osaka (JP)
- MITSUBISHI JUKOGYO KABUSHIKI KAISHA  
Tokyo (JP)

(72) Inventors:

- Miyata, Kaori, c/o Sumitomo Metal Ind. Ltd.  
Chuo-ku, Osaka-shi (JP)
- Igarashi, Masaaki, c/o Sumitomo Metal Ind. Ltd.  
Chuo-ku, Osaka-shi (JP)

- Masuyama, Fujimitsu,  
c/o Nagasaki R&DC Mitsubishi  
Nagasaki-shi, Nagasaki (JP)
- Komai, Nobuyoshi,  
c/o Nagasaki R&DC Mitsubishi  
Nagasaki-shi, Nagasaki (JP)
- Yokoyama, Tomomitsu,  
c/o Misubishi Jukogyo K.K.  
Chiyoda-ku, Tokyo (JP)

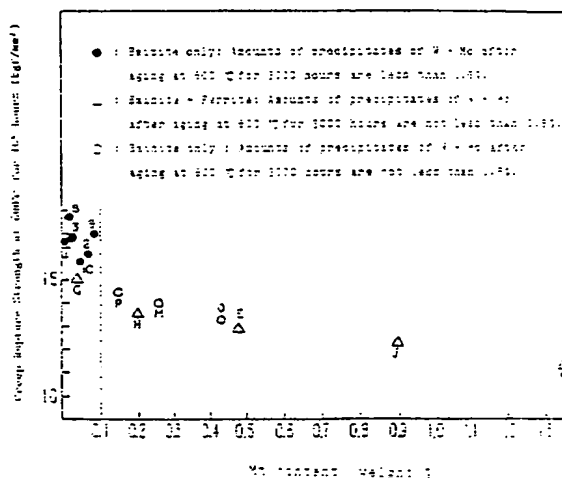
(74) Representative:

TER MEER STEINMEISTER & PARTNER GbR  
Mauerkircherstrasse 45  
81679 München (DE)

(54) A low mn-low Cr ferritic heat resistant steel excellent in strength at elevated temperatures

(57) A low Mn-low Cr ferritic heat resistant steel consisting essentially of, in weight %: 0.02-0.20% C, up to 0.7% Si, less than 0.1% Mn, up to 0.8% Ni, 0.8-3.5% Cr, 0.01-3.0% W, 0.1-0.5% V, 0.01-0.20% Nb, 0.001-0.05% Al, 0.0005-0.05% Mg, 0.0005-0.01% B, less than 0.05% N, up to 0.03% P, up to 0.015% S, 0.001-0.05% Ti and the balance Fe and incidental impurities, wherein the B content is defined so as to satisfy the following formula  $(14/11)B > N - N(V/51) / \{ (C/12) + (N/14) \} - N(Nb/93) / \{ (C/12) + (N/14) \} - N(Ti/48) / \{ (C/12) + (N/14) \}$ . The steel can further contain optionally 0.01-1.5% Mo, and/or one or more elements selected from the group consisting of 0.01-0.2% La, 0.01-0.2% Ce, 0.01-0.2% Y, 0.01-0.2% Ca, 0.01-0.2% Ta and 0.01-0.2% Zr. The steel can be used in place of the austenitic steels or high Cr ferritic steels, since it has remarkably improved toughness, workability and weldability, and excellent creep properties at elevated temperatures.

FIG. 1



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As mentioned above, the conventional method to strengthen the low Cr ferritic steel does not work sufficiently because of unstable structure, and cannot attain enough high temperature creep strength. Furthermore, the unstable structure deteriorates the toughness and other properties of the steel.

## SUMMARY OF THE INVENTION

The purpose of this invention is to provide a low Cr ferritic heat resistant steel which contains not more than 3.5% Cr, and has an improved creep rupture strength under conditions of long periods of time at high temperatures.

Another purpose of this invention is to provide a low Cr ferritic heat resistant steel which has improved toughness, workability and weldability even if it is used for thick products.

The present inventors found out the following facts A) to H) after extensive experimental work on the conditions for stabilizing the structure of the low Cr ferritic steel for long periods of time at elevated temperatures above 550°C.

A) Most of the conventional low Cr ferritic steels are Cr-Mo steels in which Mo is the main alloying element. However, W which has a larger atomic radius and a smaller diffusion coefficient than Mo can be used in larger amounts to increase the solid solution hardening effect. The large amounts of W also serves to make the precipitates stable and to improve the creep strength at elevated temperatures.

B) Fine carbides ( $M_{23}C_6$ ,  $M_7C_3$ ) containing Cr and Fe as the main elements change into coarse carbides ( $M_6C$ ) containing W, Mo and Cr as the main elements in not only the conventional Cr-Mo steel but also in the steel containing a larger amount of W after being kept at elevated temperatures above 550 °C. The coarse carbides reduce the creep strength and the toughness of the steel. Furthermore, the solid solution hardening effect of Mo and W will also be reduced because of precipitation of these elements as carbides after the steel is used for long periods of time at elevated temperatures.

C) In contrast to this, the carbides are stable and the creep strength is improved in the steel containing B even after being used at elevated temperatures for long periods. The reason is that B segregates with C so that the fine carbides,  $M_{23}C_6$ , become stable and hardly change into coarse carbides,  $M_6C$ , which reduce the high temperature strength. However, B should be added in a sufficient amount in consideration of the balance of solute B, since B tends to combine with N to form BN precipitates.

D) Large amounts of solute B are preferable to stabilize carbides, but too much B increases precipitation of  $M_{23}C_6$  carbides and makes the carbides coarse which reduces the short time creep strength and toughness. Therefore, it is preferable to reduce the amount of N and to fix solute N by Ti instead of B. Ti, as well as B, has a strong bonding force with N. However, in the reaction with C, Ti forms TiC or Ti(C,N) which precipitates with TiN as complex precipitates, although B combines with Fe, Cr and W to make  $M_{23}(C,B)_6$ , in which M means Fe, Cr and W. As mentioned above, the creep strength of the low Cr ferritic steel is controlled by the stability of  $M_{23}C_6$ ,  $M_7C_3$  and  $M_6C$ . Particularly the precipitation of coarse  $M_6C$  reduces the creep strength. Ti does not have any influence on the stability of said carbides ( $M_{23}C_6$ ,  $M_7C_3$  and  $M_6C$ ), and only has the effect to fix N.

Consequently, the solute B which satisfies the following formula (a) increases the creep strength. The formula (a) shows the balance of the B content and solute N, Ti, V and Nb contents.

$$(14/11)B > N - N(V/51) / \{(C/12) + (N/14)\} \\ - N(Nb/93) / \{(C/12) + (N/14)\} \\ - N(Ti/48) / \{(C/12) + (N/14)\} \dots (a)$$

E) Lowering Mn content serves to improve the creep strength by stabilizing  $M_{23}C_6$ , and  $M_7C_3$ , and to reduce precipitation of the coarse  $M_6C$ . The reasons are that Mn tends to precipitate with Cr and Fe as carbides and that Mn concentrated in carbides promotes coarsening of the carbides and precipitation of W.

F) As mentioned above, both B and Mn dominate stability of carbides at elevated temperatures. Therefore, the creep strength depends on the balance of B content and Mn content.

In detail, the creep strength is improved by reducing precipitation of  $M_6C$ . Reducing Mn and addition of B serve to keep fine carbides stable for long periods of time at elevated temperatures to improve the creep strength.

G) In some cases, lowering the Mn content reduces the hardenability of the steel and makes the toughness and strength lower because of formation and increase of  $\delta$ -ferrite in the steel, particularly thick steel products in which the cooling rate is low. However, the addition of B and Ti improves the hardenability, and prevents the lowering of the toughness caused by  $\delta$ -ferrite in a wide temperature range from room temperature to 550°C or higher. Furthermore, the addition of B and Ti prevents the reduction of toughness caused by coarsening of carbides.

H) In consequence, the steel structure is stabilized for long periods of time at elevated temperatures due to the complex effect of lowering the Mn content and the addition of suitable amounts of B and Ti. Accordingly, the creep

The steel of this invention is characterized in that the Mn content is suppressed to an especially low level and in that V, Nb and proper amounts of W, Ti and B are added.

Usually Mn is added to deoxidize the molten steel and to improve hot workability of the steel. However, the present inventors found the fact that Mn concentrates into carbides and reduces stability of fine carbides which serve to improve the creep strength. Particularly, if the Mn content is 0.1% or more, the transformation of fine carbides into coarse precipitates containing W, Mo and Fe as the main components ( $M_6C$  and intermetallic compounds) is accelerated when the steel is used at elevated temperatures over 550°C for long periods of time. The coarse precipitates and the precipitation of W and Mo lower the creep strength at elevated temperatures for long periods of time.

Figure 1 is a graph showing the influence of Mn content on the creep rupture strength at 600°C for  $10^4$  hours, and on amounts of precipitates of W and Mo after aging at 600°C for 3000 hours. As is apparent from Figure 1, when the Mn content is lower than 0.1%, the "amounts of precipitates of W and Mo" can be kept less than 0.5%, and the "creep rupture strength at 600°C for  $10^4$  hours" is much higher than that of the steel with Mn content not less than 0.1%.

The suppressing of Mn content is also effective to prevent precipitation of carbides around grain boundaries and coarsening of the carbides caused by addition of B. This is another reason for the improvement of the creep strength at elevated temperatures. Consequently, the Mn content is restricted to less than 0.1%.

Considering the creep rupture strength of the steel, it is desirable to lower Mn content as low as possible. However, lowering Mn content to less than 0.01% results in a very high cost of steel making under the conventional steel making process. Additionally, an extremely low Mn content reduces the hardenability of the steel, and reduces toughness in some cases when the cooling rate is small. As mentioned above, there are no lower limits of Mn content considering the creep rupture strength of the steel but it is considered that the practical target of the lower limit of Mn is 0.01%.

d) Ni:

Ni is one of the austenite stabilizing elements and improves the toughness of the steel. However, more than 0.8% Ni lowers the high temperature creep strength, and a higher content of Ni is not recommended for economical reasons. The Ni content is therefore restricted to a range of up to 0.8%.

e) Cr:

Cr is one of the indispensable elements for maintaining oxidation and corrosion resistance at high temperatures of the steel. If the Cr content is not more than 0.8%, the desired effect of Cr cannot be obtained. On the other hand, if the Cr content exceeds 3.5%, toughness, workability and thermal conductivity of the steel are lowered, and thereby advantages of the low Cr ferritic steel are reduced. The Cr content is therefore restricted in a range of 0.8 to 3.5%.

f) W:

W is effective in increasing the strength of the steel by strengthening the matrix with a solid solution of W therein and by dispersing the precipitates of the fine W carbides in the matrix. These effects of W cannot be obtained when the W content is less than 0.01%. On the other hand, the toughness, workability and weldability decrease when the W content is more than 3.0%. Accordingly, the W content is restricted in a range of 0.01 to 3.0%.

In addition, the combined addition of Mo and W is much more effective in increasing the strength, particularly creep strength, than the sole addition of W or Mo.

g) V:

V combines with the C and N to form fine precipitates of  $V(C,N)$ , which contribute to increase the creep strength at high temperatures for long periods of applied stress. If the V content is less than 0.1%, these effects cannot be fully obtained. On the other hand, if the V content is higher than 0.5%, too much precipitation of  $V(C,N)$  reduces the strength and toughness of the steel. The V content is therefore restricted in a range of 0.1 to 0.5%.

h) Nb:

As with V, Nb combines with the C and N to form fine precipitates of  $Nb(C,N)$  which contribute to increase the creep strength of the resultant steel. Particularly, Nb forms fine and stable precipitates which remarkably improve the creep strength at temperatures up to 625°C. The fine precipitate of  $Nb(C,N)$  is also effective in improving the toughness of the steel. Less than 0.01% Nb cannot achieve the above-mentioned effects, while more than 0.20% Nb increases NbC in the unsolved or precipitated state, resulting in a reduction of strength, ductility and weldability. Accordingly, the Nb content is restricted in a range of 0.01 to 0.20%.

i) Al:

Al is an essential element as a deoxidizing agent of the steel. If the Al content is lower than 0.001%, the deoxidizing effect cannot be obtained. On the other hand, more than 0.05% Al lowers the creep strength and the toughness. The Al content is therefore restricted in a range of 0.001 to 0.05%.

j) Mg:

A small amount of Mg combines with O (oxygen) and S to improve the toughness and workability of the steel. Mg is also effective to increase creep rupture ductility and strength. These effects are remarkable particularly in the steel containing V and Nb, and considerable amounts of W. If the Mg content is less than 0.0005%, the above mentioned effects cannot be obtained. On the other hand, if the steel contains more than 0.05% of Mg, not only are the effects saturated, but the workability of the steel is worsened. The Mg content is therefore restricted in a range of 0.0005 to

therein and by dispersing the precipitates of the fine carbide in the matrix. Therefore, Mo can be added optionally. The effects of Mo cannot be obtained with less than 0.01% Mo content. On the other hand, if the Mo content exceeds 1.5%, not only the effects are saturated but also the steel becomes too hard and diminishes toughness, ductility and workability. The Mo content is therefore in a range in 0.01 to 1.5%, when it is added.

p) La, Ce, Y, Ca, Ta and Zr:

These elements can be added optionally in order to control the shapes of inclusions which are formed of these elements and impurities, P, S and O. One or more of them are effective to improve the toughness, strength, workability and weldability of the steel by the above mentioned effects. However, less than 0.01% of each cannot produce these effects on the steel. On the other hand, if the alloy contains more than 0.2% of each element, the toughness and strength are worsened by excessive amounts of inclusions. Accordingly, the content of each of these elements should be in a range of 0.01 to 0.2%. When the steel contains two or more of these elements, the sum of the contents of such elements is preferably not more than 0.2%.

## EXAMPLE

Steels having the chemical compositions listed in Tables 1, 2 and 3 were melted in a vacuum melting furnace of 150 kg capacity and cast into ingots. The ingots were forged in a temperature range of 1150 to 950°C into plates of 20 mm thickness. Marks A and B refer to JIS STBA 22 and STBA 24, respectively. Both are specimens of comparative examples of the typical conventional low Cr ferritic steels.

Marks C and D are comparative examples of 2.25 Cr-1 Mo base precipitation hardening steels containing V and Nb. Marks C to K are comparative examples of steels without Ti. Marks L to P are comparative examples of steels containing various amounts of Mn. Marks Q to S are comparative examples of steels containing B and N in different ratios, and Marks T to Y are comparative examples of steels in which contents of C, Ni, Mo, V, Nb and Ti are outside of the range of this invention. The examples of the steels according to this invention are Marks 1 to 35.

Test specimens A and B were subjected to the heat treatment according to JIS, i.e., heating at 920°C for 1 hour and air-cooling. Test specimens C to S and 1 to 11 were normalized for 0.5 hour at 1050°C followed by air cooling, and then tempered for 1 hour at 780°C followed by air cooling.

After being heat-treated as mentioned above, properties of each test specimen are estimated by room temperature tensile tests, creep rupture tests and Charpy impact tests.

The room temperature tensile tests and the creep rupture tests were carried out by using test specimens of 6 mm diameter and 30 mm gauge length. The creep rupture tests were carried out at 600°C for 15,000 hours at the longest and the creep rupture strength at 600°C for 10<sup>4</sup> hours was estimated by interpolation. This creep rupture test is an accelerated test under a high stress and the results of 600°C for 10<sup>4</sup> hours guarantee the creep rupture strength at a temperature higher than 550°C for periods of time longer than 100,000 hours.

Charpy impact tests were carried out using 10 mm x 10 mm x 2 mm V-notched test specimens (JIS No. 4 specimens), and ductile-brittle transition temperatures were estimated.

Some of the specimens were subjected to aging treatment at 600°C for 3,000 hours, thereafter the specimens were dissolved in a non-aqueous solvent by the SPEED method (Selected Potentiostatic Etching by Electrolytic Dissolution Method). The extraction residue was subjected to quantitative analysis to determine contents of W and Mo in the precipitates of the specimens.

Further, in order to estimate the hardenability, ferrite phase was inspected in the specimens subjected to the heat treatment of normalizing at 1050°C for 0.5 hours and cooled with the cooling rate of 500°C/hour which is faster by 4 times than the conventional air cooling. If the steel does not have enough hardenability, ferrite phase appears after this treatment.

Test results are set forth in Tables 4, 5 and 6. The aforementioned Figure 1 shows these results arranged in order to make clear the influence of Mn content on "creep rupture strength at 600°C for 10<sup>4</sup> hours" and "amounts of precipitates of W and Mo after aging at 600°C for 3000 hours" in the examples of this invention and comparative examples.

As shown in Tables 4, 5, 6 and Figure 1, the comparative steels E, F and H to P, which contain not less than 0.1% Mn, have poor creep strength, since large amounts of coarse precipitates consisting mainly of W and Mo were formed after the long term aging.

The steel not containing Ti, such as steel G, has poor hardenability and toughness even if its Mn content is less than 0.1%.

Since the comparative steels Q to S do not satisfy the above mentioned formula (a), i.e., do not contain sufficient amounts of B, the toughness and creep strength of them are rather low because of poor hardenability.

Either toughness or creep properties are not good for the comparative steels which contain C, Ni, Mo, Mg, V, Nb and Ti in amounts outside of the range according to this invention. In these steels, too many inclusions or  $\delta$ -ferrite were formed.

The steels of this invention, as shown in Tables 5 and 6, have good ductility of more than 25% elongation. Addi-

TABLE I

Steel No.	Chemical Composition ( weight % bal.: Fe and incidental impurities )														Values of Formula (u)			
	C	Si	Mn	P	S	Ni	Cr	Mo	W	V	Nb	Ti	Al	B	N	Mg	Left Side	Right Side
A	0.12	0.37	+0.46	0.017	0.005	0.01	1.01	0.53	+	+	+	+	0.000	+	0.0139	+	10	+0.0130
B	0.11	0.30	+0.56	0.015	0.003	0.01	2.13	1.01	+	+	+	+	0.008	+	0.0158	+	10	+0.0150
C	0.13	0.17	+0.48	0.012	0.004	0.13	2.20	0.98	+	0.23	0.05	+	0.015	+	0.0078	+	10	+0.0043436
D	0.11	0.33	+0.55	0.026	0.002	0.21	2.17	0.96	+	0.21	0.06	+	0.007	+	0.0165	+	10	+0.0089036
E	0.11	0.23	+0.48	0.008	0.001	0.09	3.32	0.15	2.41	0.35	0.04	+	0.017	0.0024	0.0018	+	0.0030545	0.0003070
F	0.08	0.56	+0.48	0.005	0.011	0.08	2.56	0.09	1.71	0.17	0.08	+	0.037	0.0036	0.0078	0.002	0.0045810	0.003272
G	0.08	0.17	0.04	0.014	0.002	0.27	2.21	0.12	1.53	0.22	0.03	+	0.022	0.0049	0.0087	0.005	0.0062364	0.0031655
H	0.05	0.19	+0.20	0.013	0.003	0.29	2.24	0.07	1.47	0.21	0.06	+	0.018	0.0058	0.0080	0.003	0.0071273	0.0000733
I	0.06	0.20	+0.53	0.011	0.003	0.28	2.23	0.09	1.56	0.28	0.04	+	0.017	0.0051	0.0078	0.003	0.0064909	0.0000407
J	0.06	0.20	+0.40	0.012	0.002	0.27	2.22	0.11	1.03	0.28	0.04	+	0.016	0.0081	0.0093	0.005	0.0077636	-0.00042
K	0.07	0.18	+1.37	0.015	0.004	0.31	2.19	0.11	1.52	0.23	0.05	+	0.021	0.0048	0.0085	0.002	0.0061091	0.0018385
L	0.06	0.25	+0.65	0.009	0.004	0.30	2.25	—	1.55	0.20	0.05	0.018	0.008	0.0022	0.0077	0.005	0.0028	-0.000119
M	0.08	0.20	+0.26	0.013	0.003	0.28	2.31	0.11	1.48	0.21	0.04	0.017	0.015	0.0037	0.0068	0.004	0.0047091	0.0024763
N	0.13	0.21	+1.25	0.008	0.003	0.31	2.28	0.12	1.64	0.25	0.05	0.040	0.010	0.0041	0.0071	0.005	0.0052102	0.0036944
O	0.05	0.23	+0.43	0.012	0.002	0.29	2.24	—	1.73	0.24	0.04	0.038	0.017	0.0042	0.0083	0.004	0.0053465	-0.000657
P	0.09	0.18	+0.15	0.012	0.004	0.32	2.18	0.15	1.45	0.26	0.06	0.025	0.014	0.0050	0.0101	0.005	0.0063636	0.0030445
Q	0.18	0.22	+0.18	0.015	0.003	0.28	2.26	0.12	1.53	0.25	0.04	0.015	0.017	0.0022	0.0100	0.005	0.0028	+0.0128064
R	0.14	0.26	0.07	0.014	0.003	0.20	2.21	0.12	1.35	0.15	0.02	0.021	0.013	+0.0180	+0.0500	0.002	0.0223001	+0.0396436
S	0.13	0.25	+0.11	0.014	0.002	0.27	2.25	0.15	1.54	0.23	0.05	0.035	0.014	+	0.0082	0.005	0	+0.0045754
T	+0.30	0.10	+0.13	0.015	0.003	0.25	2.19	0.11	1.44	0.25	0.07	0.013	0.015	0.0081	0.0086	0.002	0.0103091	0.0067015
U	0.06	0.21	0.05	0.013	0.004	+1.50	2.26	1.13	2.87	0.24	0.09	0.023	0.007	0.0071	0.0180	0.005	0.0000364	0.0017528
V	0.06	0.19	0.09	0.011	0.005	0.35	2.24	+2.51	2.56	0.35	0.17	0.019	0.003	0.0050	0.0121	0.006	0.0063636	-0.005832
W	0.10	0.31	0.05	0.012	0.005	0.29	2.23	0.11	1.63	0.25	0.10	+	0.015	0.0052	0.0064	+	0.0066182	0.0039251
X	0.17	0.18	0.01	0.015	0.001	0.31	2.25	0.12	1.51	+0.85	+0.35	+	0.008	0.0095	0.0150	0.007	0.0120909	-0.005111
Y	0.06	0.20	0.06	0.009	0.003	0.25	2.23	0.35	1.62	0.20	0.08	+0.075	0.015	0.0085	0.0081	0.007	0.0108182	0.0011596

Note: Mark "+" indicates the value outside of the range according to the present invention.

TABLE 8

Steel No.	Chemical Composition ( weight % bal.: Fe and Incidental Impurities )																	Values of Formula (n)	
	C	Si	Mn	P	S	Ni	Cr	Mo	W	V	Nb	Ti	Al	O	N	Hg	Others	Left Side	Right Side
21	0.071	0.54	0.01	0.005	0.003	0.04	2.54	0.28	1.60	0.18	0.07	0.042	0.003	0.0038	0.0152	0.002	La:0.02,Ca:0.01, Ce:0.02, Y:0.02, Ta:0.02	0.0040364	0.0040055
22	0.075	0.09	0.01	0.015	0.003	0.29	2.24	—	1.69	0.23	0.04	0.030	0.028	0.0055	0.0093	0.005	La:0.01,Ca:0.01, Ce:0.01	0.007	0.001815
23	0.072	0.18	0.03	0.015	0.003	0.31	2.27	—	1.67	0.19	0.04	0.025	0.019	0.0048	0.0089	0.003	La:0.02,Ca:0.04, Ce:0.02, Y:0.02, Ta:0.02	0.0061091	0.0026278
24	0.066	0.07	0.04	0.013	0.002	0.27	2.10	0.13	1.61	0.22	0.06	0.016	0.021	0.0050	0.0086	0.002	La:0.04, Y:0.02, Ta:0.02	0.0063630	0.001856
25	0.054	0.16	0.07	0.013	0.004	0.27	2.23	0.09	1.62	0.24	0.05	0.012	0.025	0.0080	0.0105	0.002	La:0.04,Ca:0.04, Ce:0.02, Y:0.02, Ta:0.01	0.0101010	-0.000487
26	0.052	0.20	0.02	0.011	0.002	0.25	2.31	0.14	1.68	0.19	0.05	0.033	0.017	0.0018	0.0120	0.015	—	0.0022909	0.0005545
27	0.063	0.19	0.05	0.013	0.003	0.33	2.28	0.13	1.64	0.21	0.05	0.015	0.019	0.0061	0.0160	0.001	La:0.01,Zr:0.02, Y:0.01	0.0077636	0.003212
28	0.071	0.20	0.04	0.013	0.002	0.31	2.29	0.12	1.73	0.22	0.05	0.011	0.021	0.0032	0.0101	0.006	—	0.0040727	0.0023699
29	0.140	0.15	0.07	0.011	0.002	0.22	2.28	1.01	2.01	0.26	0.05	0.025	0.020	0.0070	0.0110	0.005	—	0.0100545	0.0057348
30	0.062	0.18	0.03	0.012	0.002	0.32	2.26	0.65	1.59	0.26	0.05	0.024	0.019	0.0053	0.0101	0.020	—	0.0067455	-0.000425
31	0.060	0.10	0.07	0.013	0.004	0.27	2.23	0.25	1.62	0.24	0.05	0.018	0.025	0.0060	0.0105	0.001	—	0.0076364	0.0002401
32	0.030	0.15	0.02	0.012	0.002	0.15	0.98	1.05	1.01	0.23	0.02	0.011	0.020	0.0035	0.0041	0.002	Ca:0.01, Y:0.01	0.0044545	-0.003173
33	0.150	0.16	0.02	0.005	0.002	0.11	2.24	0.13	1.71	0.40	0.02	0.032	0.012	0.0070	0.0150	0.007	—	0.0080091	0.0053567
34	0.050	0.60	0.01	0.015	0.010	0.12	2.37	—	1.63	0.25	0.05	0.025	0.025	0.0075	0.0065	0.030	—	0.0095455	-0.001866
35	0.030	0.02	0.02	0.013	0.002	0.13	2.51	—	1.32	0.12	0.17	0.020	0.017	0.0050	0.0093	0.002	Ta:0.01,Zr:0.01	0.0063636	0.0040629

TABLE 5

Steel No.	Room Temperature Tensile Test			Ductile-Brittle Transition Temp. in Charpy Impact Test (°C)	600 °C, 10 <sup>4</sup> hours Creep Rupture Strength (kgf/mm <sup>2</sup> )	Amounts of Wit-Mo Precipitates after 600 °C, 3000 hours Aging (weight %)	Hardenability
	Tensile Strength (kgf/mm <sup>2</sup> )	0.2% Proof Strength (kgf/mm <sup>2</sup> )	Elongation (%)				
1	66.3	54.3	27.1	-32	15.9	—	◎
2	66.7	54.9	29.5	-39	16.1	0.35	◎
3	67.1	55.1	26.8	-35	16.8	0.34	◎
4	65.4	53.0	30.8	-41	16.2	—	◎
5	65.6	53.6	28.4	-23	10.1	0.35	◎
6	66.1	53.8	20.1	-36	10.0	—	◎
7	63.5	52.8	33.3	-30	16.7	0.33	◎
8	68.3	57.3	28.0	-41	17.7	0.23	◎
9	70.5	57.9	27.4	-28	17.0	0.38	◎
10	62.3	53.4	38.1	-41	15.8	0.29	◎
11	71.0	59.8	20.5	-33	17.0	—	◎
12	63.4	53.7	34.1	-41	15.7	—	◎
13	68.7	57.4	26.5	-41	10.8	—	◎
14	69.2	57.8	27.8	-45	17.8	—	◎
15	66.1	55.3	28.1	-43	17.1	—	◎
16	67.2	56.7	27.4	-45	16.4	—	◎
17	61.2	51.8	32.4	-40	17.7	—	◎
18	66.4	54.5	33.8	-50	15.8	—	◎
19	64.7	64.5	30.1	-51	16.2	—	◎
20	66.1	55.1	28.9	-45	16.0	—	◎

Note: In the column of Hardenability, ◎ : Bainite only (good hardenability)

0.02-0.20% C. up to 0.8% Ni. 0.1-0.5% V. 0.0005-0.05% Mg. up to 0.03% P.	up to 0.7% Si 0.8-3.5 % Cr. 0.01-0.20% Nb. 0.0005-0.01% B. up to 0.015% S.	less than 0.1% Mn. 0.01-3.0% W. 0.001-0.05% Al. less than 0.05% N. 0.001-0.05% Ti.
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optionally one or more elements selected from the group consisting of:

0.01- 0.2% La. 0.01- 0.2% Ca.	0.01-0.2% Ce. 0.01-0.2% Ta	and	0.01-0.2% Y, 0.01-0.2% Zr.
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and the balance Fe and incidental impurities, wherein the B content is defined so as to satisfy the following formula:

$$\begin{aligned}
 (14/11)B > N - N(V/51) / \{(C/12) + (N/14)\} \\
 - N(Nb/93) / \{(C/12) + (N/14)\} \\
 - N(Ti/48) / \{(C/12) + (N/14)\}.
 \end{aligned}$$

2. A low Mn-low Cr ferritic heat resistant steel excellent in high temperature strength consisting essentially of, in weight %:

0.02-0.20% C. up to 0.8% Ni. 0.01-1.5% Mo. 0.001-0.05% Al. less than 0.05% N. 0.001-0.05% Ti.	up to 0.7% Si. 0.8-3.5 % Cr. 0.1-0.5% V. 0.0005-0.05% Mg. up to 0.03% P.	less than 0.1% Mn. 0.01-3.0% W. 0.01-0.20% Nb. 0.0005-0.01% B. up to 0.015% S.
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optionally one or more elements selected from the group consisting of:

0.01- 0.2% La. 0.01- 0.2% Ca.	0.01-0.2% Ce 0.01-0.2% Ta	and	0.01-0.2% Y, 0.01-0.2% Zr.
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and the balance Fe and incidental impurities, wherein the B content is defined so as to satisfy the following formula:

$$\begin{aligned}
 (14/11)B > N - N(V/51) / \{(C/12) + (N/14)\} \\
 - N(Nb/93) / \{(C/12) + (N/14)\} \\
 - N(Ti/48) / \{(C/12) + (N/14)\}.
 \end{aligned}$$





European Patent  
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## EUROPEAN SEARCH REPORT

Application Number  
EP 97 10 1998

DOCUMENTS CONSIDERED TO BE RELEVANT			
Category	Citation of document with indication, where appropriate, of relevant passages	Relevant to claim	CLASSIFICATION OF THE APPLICATION (Int.Cl.6)
A	EP 0 505 732 A (SUMITOMO METAL IND ;MITSUBISHI HEAVY IND LTD (JP)) 30 September 1992 * claim 1 *	1,2	C22C38/22 C22C38/24 C22C38/26 C22C38/28 C22C38/32
D,A	--- PATENT ABSTRACTS OF JAPAN vol. 014, no. 522 (C-0778), 15 November 1990 & JP 02 217438 A (SUMITOMO METAL IND LTD), 30 August 1990, * abstract *	1,2	
D,A	--- PATENT ABSTRACTS OF JAPAN vol. 012, no. 223 (C-507), 24 June 1988 & JP 63 018038 A (KAWASAKI STEEL CORP;OTHERS: 01), 25 January 1988, * abstract *	1,2	
D,A	--- PATENT ABSTRACTS OF JAPAN vol. 012, no. 283 (C-518), 3 August 1988 & JP 63 062848 A (MITSUBISHI HEAVY IND LTD), 19 March 1988, * abstract *	1,2	
A	--- EP 0 560 375 A (SUMITOMO METAL IND ;MITSUBISHI HEAVY IND LTD (JP)) 15 September 1993 * claim 1 *	1,2	
The present search report has been drawn up for all claims			
Place of search THE HAGUE		Date of completion of the search 7 May 1997	Examiner Gregg, N
CATEGORY OF CITED DOCUMENTS		I : theory or principle underlying the invention E : earlier patent document, but published on, or after the filing date D : document cited in the application I : document cited for other reasons & : member of the same patent family, corresponding document	
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